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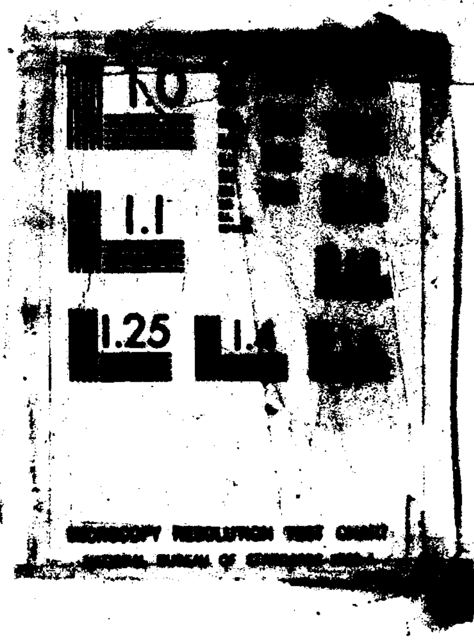
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DEFORMATION AND FRACTURE OF P/M TITANIUM ALLOYS

DONALD A. KOSS
DEPARTMENT OF METALLURGICAL ENGINEERING
MICHIGAN TECHNOLOGICAL UNIVERSITY
HOUGHTON, MI 49931

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Progress is reviewed for a research program whose primary purpose has been to provide a broad-based understanding of the deformation and fracture of high performance alloys in general and titanium alloys in particular. The research has ranged from strengthening mechanisms in titanium alloys to fundamental studies of crack propagation, localized necking, hydrogen embrittlement, and ductile fracture utilizing engineering alloy behavior.

(Continued)

20. Abstract (continued)

Progress for the period 1 July 1975 to 1 October 1984 is reviewed for the following portions of this research program:

(1) the influence of silicon on the age hardening of beta-phase Ti alloys,
(2) the flow and fracture behavior of multi-phase alloys with lamellar microstructures,
(3) fracture along planar slip bands,
(4) the multiaxial deformation and fracture of plastically anisotropic alloys in the form of sheet,
(5) the influence of stress state on the hydrogen embrittlement of hydride-forming alloys in general and Ti in particular,

and (6) the deformation and fracture of alloys containing pores and voids. Keywords;

(19)

INTRODUCTION

Titanium-base alloys have been used extensively in applications requiring high strength to weight ratios, good fracture resistance, as well as resistance to environmental degradation. In addition to being subjected to stringent mechanical behavior demands, Ti alloys are also limited in use by their high cost, especially in parts of complex shape where considerable material waste often occurs and extensive processing is required. This has led to the application of advanced processing techniques, such as powder metallurgy (PM), to high performance alloys in general and Ti alloys in particular. The primary purpose of this research program has been to provide a broad-based understanding of the deformation and fracture behavior of high performance alloys in general and titanium alloys, including PM alloys, in particular. The research has ranged from experimental studies of strengthening mechanisms in Ti alloys to theoretical analyses of fracture and crack growth phenomena. As such, much of the research should also be viewed as fundamental studies of deformation and fracture of high performance alloys, utilizing Ti alloys as model systems. The present report summarizes progress in the several areas of research for the period 7/1/75 to 10/1/84 performed under the auspices of Contract No. N00014-76-C-0037. These areas may be roughly grouped by the following (sequential) themes:

1) the influence of silicon on the age hardening of beta-phase Ti alloys,
2) the flow and fracture behavior of multi-phase alloys with lamellar microstructures,
3) fracture along planar slip bands,
4) the multiaxial deformation and fracture of plastically anisotropic alloys in the form of sheet

5) the influence of stress state on the hydrogen embrittlement of hydride-forming alloys in general and Ti in particular, and

6) the deformation and fracture of alloys containing pores and voids.

An important aspect of this program is the educational experience that it has provided to the graduate students involved. The following graduate students have been supported by this program during the past ten years: (1) Steven Tuominen (Ph.D.), (2) Donald Graham (M.S.), (3) Kwai Chan (M.S., Ph.D.), (4) Roy Bourcier (M.S., Ph.D.), (5) Charles Lentz (M.S.), (6) Barbara Lograsso (M.S., currently Ph.D. candidate), (7) Paul Magnusen (M.S., currently Ph.D. candidate), (8) Dale Gerard (M.S., currently Ph.D. candidate) and (9) Ellen Dubensky (M.S. candidate). It may be also noted that the research performed has resulted in twenty-four technical reports and twenty-six publications (not counting those currently under review).

SUMMARY OF RESEARCH

1. **The Influence of Silicon on the Age Hardening of Beta-phase Ti Alloys.**

(S. Tuominen, Ph.D., G. Franti, postdoctoral research associate, and D. Graham, M.S.)

Commercial β-phase Ti alloys are strengthened primarily by solid solution hardening of the β-phase matrix and age-hardening by α-phase and/or ω-phase formation. An alternate means of strengthening β-phase Ti alloys is to use Si-induced precipitation hardening. As a continuation of earlier studies[1,2], this program examined the influence of Si on the microstructure and properties of both stable and metastable β Ti-V alloys. In a Ti-40V-1Si alloy with a relatively stable bcc β-phase matrix, the following is the precipitation sequence which occurs upon aging quenched samples: Bcc supersaturated solid solution → Bcc zones → $(Ti,V)_x\ Si_y$ (hexagonal) → $(Ti,V)_3$ Si (Tetragonal)[3]. The formation of both the bcc zones and the hexagonal precipitates result in hardness peaks. However, because of the structure and morphology of these two types of precipitates, they are sheared during deformation; the work hardening of the alloy thus remains low and tensile ductility is limited[2].

In an attempt to improve the strength/ductility relationship in β Ti alloys, the multiphase strengthening of a Ti-30V-1Si alloys by the formation of both silicides and the hcp α-phase has also been studied[4]. Structure-property relationships have been examined in materials aged to form only silicides as well as both silicides and the α-phase. The optimum combination of strength and tensile ductility was obtained upon aging as-quenched material at 450°C by using prior cold work to promote uniform nucleation and growth of both the silicides and an α-phase which is solid solution hardened by Si.

2. **The Flow and Fracture Behavior of Multi-phase Alloys with Lamellar Microstructures** (Kwai Chan, M.S. and Ph.D.)

Many alloys possess a microstructure which contains colonies comprised of a mixture of phases with a lamellar morphology. Some examples are the α-β alloys in the β-annealed condition, steels containing pearlite, and some directionally solidified eutectics. While certain features of the deformation of such alloys are well established, relatively little is known about the deformation behavior of individual colonies. It is well known that some, but not all, colonies with a lamellar microstructure are susceptible to very planar, nonuniform slip which can lead to easy crack initiation and propagation along such localized bands. This is particularly so in α-β Ti alloys[5-10] and pearlitic steels[11-16]. Even though it may be confined to a few isolated colonies, such behavior can limit the overall ductility and fatigue resistance of a component. There has been no experimental study which identifies the relationships between a microstructure consisting of aligned lamellae of two deformable phases and the resulting deformation behavior for individual colonies. Utilizing the α-β Ti alloy, Ti-8Al-1Mo-1V, this investigation examined the deformation and fracture behavior of individual colony test samples[17]. The yield strength, work hardening behavior, fracture toughness and the slip characteristics have been related to several microstructural parameters, such as the orientation of the colony to the stress axis and the crystallography of slip. The Ti-8Al-1Mo-1V alloy was used as a model system because it is relatively well characterized; the nature of the individual phases can be manipulated, and it is much easier to grow rather large α-β Widmanstatten colonies in this alloy than in other α-β alloys, such as Ti-6Al-4V. In this system, the titanium alloy microstructure consists of alternating platlets of the two phases, both of which are ductile and whose elastic modulii, yield strengths, and work hardening exponents [as

deduced from polycrystalline material of similar compositions to the individual phases] are roughly within 25% of each other[18].

The influence of the alignment of the alpha and beta phases on the yielding and flow behavior of single-crystal type specimens has been examined for single Widmanstatten colonies of the Ti-8Al-1Mo-1V alloy. The results may be summarized as follows[17]:

(a) there is a large (>2X) variation in the critical resolved shear stress for yielding of individual colonies [the only exception is slip within the alpha phase parallel to the beta phase],

(b) the active macroscopic slip system is often not that system subjected to the highest resolved shear stress,

(c) colonies with a high resolved shear stress for yielding also exhibit a high work hardening rate and fine, uniform slip,

(d) the above behavior appears to be independent of which slip system [basal, prism, or pyramidal] is active in the alpha phase, and

(e) correlating the observed results with several orientation parameters indicates that the yield stress increases as the slip direction of the active slip system approaches normality to the interphase boundary.

The above results suggest several conclusions regarding the yielding and flow behavior of alloys with aligned, lamellar phase mixtures. First, an inspection of the data regarding the presence of prism and pyramidal slip as well as basal slip indicate that the crystallographic alignment of a slip system in the softer alpha phase with a potential slip system in the beta-phase lamellar does not cause preferred slip on the "aligned" system. Thus, a model based on an alignment of slip systems between the phases would be inadequate as a basis for predicting yielding and flow behavior. Secondly,

the data also indicates that elastic interaction stresses at the phase interfaces[19] are not a principal factor in controlling macroscopic deformation in the alpha-beta colonies examined. Third and most important, the data provide strong evidence that the macroscopic flow behavior of individual grains or colonies comprised of aligned, ductile lamellae depends on the ability of a slip system which is activated in the softer phase to shear through the harder phase.

The concept that slip may be activated in the softer phase but not penetrate the harder phase has significant implications. In such an instance, an additional stress is required to activate a new, more "potent" slip system. As a result, a high yield stress is usually associated with slip on systems with a comparatively low shear stress but with a high "shearing" ability[17,20]. The presence of more highly stressed slip systems also implies their activity in the softer phase. Thus a high yield stress associated with uniform, dispersed slip results and, given the multiplicity of slip within the softer phase, is accompanied by a higher rate of work hardening. In contrast, if the first slip system activated in the softer phase can shear the harder phase, yielding at a low macroscopic stress occurs and is characterized by planar slip, low hardening, and easy crack initiation.

In summary, the experimental data clearly indicate that the alignment of phases has a pronounced effect on the resulting yielding and flow behavior even if both phases are ductile and do not differ greatly in elastic and plastic properties. An obvious microscopic yielding and flow criterion for such materials is that macroscopic yielding occurs only when a slip system, once activated in the softer phase, shears the harder phase. Such criterion has several implications as regards yielding and flow behavior especially if the most highly stressed slip system may not be able to shear the harder

phase; some of the implications have been discussed above. The problem is that, despite our attempts, there exists to date no theoretical analysis capable of predicting the "shearing ability" of a specific slip system under a given set of experimental conditions. Thus, our understanding as well as our ability to predict the deformation behavior of alloys with lamellar microstructures remain incomplete.

The fracture toughness behavior of individual α-β Widmanstatten colonies of the Ti-8Al-1Mo-1V alloy in sheet form was also investigated[21]. Crack extension occurs predominantly across the α-β lamellae under conditions of plane stress and on planes which are inclined to both the thickness and the width of specimens. Crack tip plasticity in both single-colony and polycrystalline material is dominated by through-thickness deformation involving slip and often twinning. In the single-colony specimens the crack tip plasticity was characterized experimentally by identifying the active slip/twinning planes and by calculating the distribution of shear stresses on the possible deformation systems using shear criteria as defined on both microscopic and macroscopic scales. As suggested by the fracture behavior of the polycrystalline specimens, the fracture plane across individual colonies is near that slip or twinning plane that experiences the largest shear stress when the shear stress is a maximum on a macroscopic scale. The plane stress fracture toughness of individual colonies depends on colony orientation and on the nature of the deformation at the crack tip. High toughness of a colony is associated with multiple slip and twinning and with the absence of low energy fracture along or near interfaces such as twin boundaries of α-β interfaces. Conversely, slip localization into coarse basal slip bands coplanar with the crack results in the extension of the crack along such bands and thus in the lowest fracture toughness observed. Such fracture behavior can be readily

understood in terms of an elastic-plastic model for crack advance along slip bands coplanar with a crack[22,23].

3. Fracture Along Planar Slip Bands (Kwai Chan, Ph.D., and James Wilcox, M.S.)

Fracture along planar, inhomogeneous slip bands is a common occurrence in high strength alloys. Such fracture behavior can occur under both tensile and cyclic loading conditions and has been reported in fcc, bcc, as well as hcp alloys. For example, tensile fracture along planar slip bands has been observed in Al-[24], Ni-[25,26], β (bcc) Ti-[21], and α (hcp)-β Ti[26,27] alloys. Crack growth along planar slip bands under cyclic loading, e.g., crystallographic Stage I fatigue is also very common to a large number of high strength, structural alloys: Al-base[28], Ni-base[29-31], brass[32], and α-β Ti[33-37] alloys. In all cases, the crack path occurs along planes which are subject to large shear stresses and shear strains. At the same time, this "slip band decohesion" process is usually characterized by a cleavage-like or a substantially brittle appearance. Thus, many in investigators have concluded that normal stresses are also important in slip-band fracture processes. An important aspect of this research program has been the development and application of a straight-forward theoretical analysis of cracking along planar slip bands[22,23]. Under such conditions, the resulting state of stress near the crack-tip is conducive both for continued localization of planar slip ahead of the crack and for the development of large normal stresses near the crack-tip. Many of the features in the fracture/fatigue studies noted above can be readily understood in terms of these concepts.

The basis for the analysis is the simple model in which a mixed mode crack propagates along a slip band which is coplanar with the crack and whose slip vector is also confined to the crack plane. The result is that although

the planar slip band extends to r_p ahead of the crack, the normal stresses remain elastic until non-coplanar, secondary slip is activated at r_p^s close to the crack-tip. Calculations[22,23] show that: (a) once a crack with a coplanar slip band has formed, activating secondary slip (with a shear displacement component perpendicular to the crack plane) is difficult, (b) the difficulty of activating secondary slip in turn results in large normal stresses near the crack-tip and (c) the stress state is sensitive to the crystallography of slip with the largest normal stresses developed for basal plane cracking in a hcp alloy. The combination of inhomogeneous shear and large normal stresses results in easy crack propagation and low toughness usually, but not always, associated with inhomogeneous shear band cracking. Any work softening, such as cyclic softening which might occur in the planar slip band will serve to accentuate these effects and crystallographic Stage I fatigue crack propagation would be a natural consequence.

The propagation of an inclined crack along a coplanar slip band was also examined with regard to the conditions usually associated with hydrogen embrittlement[38]. The analysis shows that large normal stresses are located very near to the crack-tip, the presence of the planar slip at the crack-tip, and a small but probably significant amount of non-coplanar slip all contribute to a set of conditions which are very favorable to hydrogen embrittlement. In view of the analysis, the effect of gaseous environment on the fatigue crack propagation behavior of an age hardenable metastable β-phase Ti-30V alloy was examined[38]. Tests on both single crystal and polycrystalline specimens show that, consistent with the analysis, gaseous hydrogen accelerates the fatigue crack growth rates <u>only</u> when Stage I fatigue occurs along a crystallographic {112} slip plane.

4. <u>The Multiaxial Deformation and Fracture of plastically Anisotropic Alloys in the Form of Sheet</u> (Kwai Chan, Ph.D., and Charles Lentz, M.S., Prof. K. Weinmann, Michigan Tech, and Dr. A. K. Ghosh, Rockwell International Science Center)

Titanium alloys in sheet form usually possess crystallographic textures which, in some instances, can be quite strong. Due to the nature of slip in the hcp α-phase, strong textures in α-β Ti alloys often exert a pronounced influence on the mechanical properties. Many of these effects are well documented for an alloy such as Ti-6Al-4V in the form of plate and bar stock. In the case of sheet metal deformation, a crystallographic texture usually results in plastic anisotropy, the degree of which can be measured quantitatively by the plastic anisotropy parameter R which is the ratio of the width strain to thickness strain in a uniaxial tensile test. Previous studies of the deformation of strongly textured Ti alloy sheet have been primarily confined to multiaxial yielding behavior and texture strengthening (at small strains) as influenced by the very wide range of R-values (0.2 to 14) possible in these alloys[39-42]. Large strain deformation and Ti and Ti-6Al-4V sheet has been examined with regard to the effect of strain hardening and strain-rate hardening on necking behavior in uniaxial tension[43] and forming limit strains in stretch forming behavior[44,45]. However, the plastic anisotropy was not varied in these studies, and no attempt has been made to determine the influence of R on large strain deformation of Ti and its alloys.

Large strain deformation and the phenomenon on localized necking in sheet metal has been studied in considerable detail in steel, brass and aluminum alloys (see, for example, References 46 and 47). It is well established in these materials that strain hardening and strain-rate hardening are both

important in enhancing the resistance to localized necking and thus the formability. The influence of crystallographic texture and R-value on the stretch formability is, however, less conclusive[47-50]. The experimental difficulty has been in manipulating sheet metal processing to vary R over a large range of values without changing other properties such as strain hardening exponent, n, and the strain-rate sensitivity exponent, m. In addition, owing to the nature of slip in fcc and bcc metals, the range of R-values studies is not large; usually $0.5 \leq R \leq 2$. It is therefore difficult to separate the effect of crystallographic texture and R value from that of n and m. Thus, our knowledge of the influence of plastic anisotropy on the large strain deformation of sheet material in general and Ti alloys in particular is limited by the nature of the experimental studies to date.

The influence of crystallographic texture on the deformation and fracture behavior of strongly textured Ti alloy sheet has been investigated in both uniaxial and multi-axial tension[51-53]. Uniaxial tensile tests have been performed on Ti-6Al-4V and Ti-5Al-2.5 Sn sheets with both a basal and a basal-transverse texture and R-values ranging from 0.5 to 12[52]. The results indicated that, by controlling the ease of through-thickness slip, the crystallographic texture strongly affects the plastic anisotropy of the material but has relatively little effect on the strain-rate sensitivity and work-hardening rates at large strains. A strong resistance to through-thickness slip, manifested by a high R-value, enhances the post-uniform elongation and the ability of the material to retain the load-carrying capacity beyond maximum load in uniaxial tension. This behavior can be qualitatively understood in terms of the effect of R on the hardening which occurs as the strain state within the diffuse neck shifts from uniaxial tension toward plane strain. A higher R-value also increases significantly the limit strain at the onset of localized necking as well as the fracture

strain. The effects of R-value on the limit strain can be qualitatively understood in terms of a critical thickness strain criterion and can be quantitatively predicted by two analyses, one of which assumes an imperfection to be present while the other does not.

In multiaxial stretch-forming operations the effects of work hardening and strain rate hardening clearly dominate, and plastic anisotropy is usually considered to be a minor factor[47] (see References 47 and 48 for reviews). This is understandable for fcc and bcc alloys where minor changes of the strain or strain rate hardening can dominate the effects of plastic anisotropy because the anisotropy caused by slip in fcc/bcc metals is relatively small. As a result, the interpretations of the existing data which relate stretch forming and plastic anisotropy has been conflictive[47-50]. On the other hand, the effects of plastic anisotropy on large strain deformation can be quite pronounced in strongly textured hcp alloys. Using the R-value as a measure of plastic anisotropy, the influence of crystallographic texture and R-value on the multiaxial stretch forming behavior of strongly textured Ti alloy sheet has been examined[51,53]. The study was based on strongly textured Ti-6Al-4V sheet with R-values ranging from 0.5 to 12 but with relatively constant strain and strain-rate hardening exponents. The results indicated that a high R-value and difficult through-thickness slip enhance the forming limit as well as fracture strains when the minor strain in the plane of the sheet is negative, this effect being most pronounced at uniaxial tension. At plane strain, the R-value has little or no influence on the limit or fracture strain. A direct determination of the effect of R-value on the biaxial stretch forming characteristics of Ti-6-4 sheet is precluded by the intervention of fracture prior to localized necking when the minor strain is

positive. The influence of plastic anisotropy on both the localized necking and the fracture behavior can be generally understood in terms of the difficulty of attaining critical thickness strains as through-thickness slip becomes more difficult.

Sheet materials deforming under multiaxial states of stress, as in sheet metal forming operations, usually fail by localized necking. The current interest in understanding sheet metal formability has led to several theoretical analyses of localized necking based on different criteria. These localized necking criteria include: a localized shear zone along a direction of zero-extension[54], materials imperfections[55], the presence of a vertex on the yield surface[56], and void growth[57].

Localized necking along a direction of zero-extension was originally proposed by Hill[54]. Hill's theory predicts that the maximum principal strain ϵ_1^* prior to localized necking (i.e., the limit strain) has a magnitude of $\epsilon_1^* =$ n at plane strain and increases to $\epsilon_1^* = (1 + R)\, n$ for the uniaxial tension deformation of sheet exhibiting normal anisotropy with a plastic anisotropy parameter R, which is defined as the ratio of the width strain to thickness strain of sheet specimens deformed under uniaxial tension. For plastically isotropic material (R = 1), the limit strain at uniaxial tension reduced to the well-known $\epsilon_1^* = 2n$ expression. Hill's theory, however, does not take into account the strain-rate hardening of the material or preexisting imperfections, and it cannot explain the phenomenon of localized necking under biaxial stretching.

Strain localization developed by local weakness of material (imperfection) was first proposed by Marciniak and Kuczynski (M-K)[55] and extended by Sowerby and Duncan[58] as a means of describing localized necking in

biaxial stretching when the minor principal strain ε_2 is positive. The M-K analysis assumes the presence of a material imperfection in the form of a groove or trough. Imposing the same ε_2 inside and outside the groove while proportional straining is maintained outside the groove. M-K have shown that deformation within the groove occurs at a faster rate than the rest of the sheet. The concentration of strain (ε_1) within the groove eventually leads to the plane strain condition ($d\varepsilon_2 = 0$) within the groove and to localized necking. The M-K model is thus able to include the effects of strain rate hardening and to explain localized necking in biaxial stretching. It has been used with reasonable success to calculate the biaxial stretch formability of A-K steel and 70-30 brass[59].

As part of this program, the localized necking in sheet metal has been examined for strain paths between uniaxial tension and plane strain (i.e., the negative minor strain region of a forming limit diagram)[60]. The behavior of sheet with preexisting imperfections has been analyzed (extending the M-K theory[55]) and is contrasted to that free of imperfections (based on the Hill theory[54]. In particular, it is shown that the size and orientation of an imperfection is critical in determining whether or not localized necking is initiated along the imperfection. The influence of strain hardening, strain-rate hardening, and plastic anisotropy on localized necking of an imperfect sheet is also examined. One of the most significant conclusions obtained from present analysis and from a reexamination of Hill's theory is the prediction of a critical thickness strain criterion for the onset of localized necking at negative minor strains, regardless of whether or not an imperfection is present. The critical thickness strain criterion is observed in Ti alloys, Al alloys, steels, and brass.

5. <u>The Influence of Stress State on the Hydrogen Embrittlement of Ti and Ti Alloys</u> (Roy Bourcier, Ph.D., Charles Lentz, M. S., Barbara Lograsso, M.S., and Dale Gerard, M.S., Drs. M. G.Stout and S. S. Hecker, Los Alamos National Laboratory)

Although commercially pure (C.P.) titanium can be embrittled by hydrogen, the degree of embrittlement has been observed to depend on many factors. For example, smooth bar specimens tested at slow strain rates in uniaxial tension usually show relatively few signs of embrittlement at levels of hydrogen up to 400 wt ppm[61-63]. On the other hand, substantial embrittlement at similar hydrogen contents occurs in bar or plate material upon the introduction of notches, decreasing the test temperature, and increasing the strain-rate, oxygen content, or grain size[61-66]. Metallographic evident indicates that these embrittlement effects are associated with the introduction of cracks into the titanium matrix by the local fracture of titanium hydride precipitates[62,63].

In recent years, there has been an increasing demand for C.P. Ti sheet or tubing products. Although the sheet and tubing typically are loaded in service under multiaxial states of stress and will deform under conditions of plane stress, the hydrogen embrittlement of C.P. Ti under these conditions has not been characterized. The previous results on notched tensile or impact specimens indicate that embrittlement can occur in thick section components under conditions of a large hydrostatic stress[61,64-66]. However, such a stress state is local to the tip of the notch, difficult to characterize in a deforming body, and has a degree of triaxiality (as measured by the hydrostatic stress/equivalent shear stress ratio) which is four to five times greater than that possible even in the equibiaxial tension of smooth sheet or thin wall tubing.

Utilizing the techniques of punch-stretch testing applied to hydrogen embrittlement[67], the embrittlement of C.P. Ti sheet at three levels of hydrogen content (60, 630 and 980 wt ppm) and deformed over a range of multiaxial stress states from uniaxial tension to balanced biaxial tension has been studied[68]. The data show that hydrogen embrittlement of plastically anisotropic Ti sheet depends on stress state, being the most severe in equibiaxial tension. Quantitative metallography indicates that the effect of stress state is primarily a result of two factors: (1) plane strain and equibiaxial tensile deformation are especially effective in causing the strain-induced fracture of hydrides and consequently void formation, and (2) the void link-up process in plane strain and equibiaxial tension initiates at a comparatively low bulk void density. The results are analyzed in terms of the influence of stress state on both hydride fracture and the occurrence of shear instabilities triggered by hydride fracture/void nucleation.

The influence of internal hydrogen on the multiaxial stress-strain behavior of commercially pure titanium has also been studied[69]. Thin-walled specimens containing either 20 or 1070 ppm hydrogen have been tested at constant stress ratios in combined tension and internal pressure. The addition of hydrogen lowers the yield strength for all loading paths but has no significant effect on the strain hardening behavior at strains $\varepsilon > 0.02$. Thus, the hydrogen embrittlement of titanium under plain strain or equibiaxial loading is not a consequence of changes of flow behavior. The yielding behavior of this anisotropic material is described well by Hill's quadratic yield criterion. As measured mechanically and by pole figure analysis, the plastic anisotropy changes with deformation in a manner which depends on stress state. Hill's criterion and the associated flow rule do not describe the stress-strain behavior well because of their inability to account for changes of texture which depend on multiaxial stress path. Hence, a strain

dependent, texture-induced strengthening effect in equibiaxial tension is observed, this effect having the form of an enhanced strain hardening rate.

The studies described above have shown that commercially pure (α-phase) Ti sheet exhibits a dramatic loss of ductility due to internal hydrogen under balanced biaxial deformation even though no loss of ductility is detected in uniaxial tension. However, α-β Ti alloys do not readily exhibit hydrogen embrittlement expect at very high hydrogen contents or in the presence of locally high triaxial stresses caused by cracks or notches. This behavior may be interpreted to result from the high solubility of hydrogen in the bcc β-phase which acts as a sink for the hydrogen in α-β Ti alloys. Thus the role of the crack or notch is to provide the local triaxial stress state which acts to accumulate hydrogen to the level necessary for hydride formation. The following question remains: will an α-β or β-phase Ti alloy be susceptible to hydrogen embrittlement under uniform multiaxial loading (no cracks or notches) if the hydrogen remains in solution (hydrides do not form)?

The influence of hydrogen on the deformation and fracture behavior of β-phase Ti-30V and α-β Ti-6Al-4V alloy sheet under multiaxial states of stress has thus also been examined[70]. Uniaxial tensile and punch-stretch tests have been utilized to examine Ti-30V containing 40 or 2000 wt. ppm H as well as Ti-6Al-4V with either 30, 250, or 500 wt. ppm H. The behavior of Ti-30V in uniaxial tension at room temperature indicates that at levels of 2000 wt. ppm, H has no effect on the strength, but it causes a small increase in strain-rate hardening and a small decrease in strain hardening. The net effect is that <u>no</u> significant change in either the localized necking or fracture behavior for any loading path between uniaxial tension and balanced biaxial stretching.

Similarly the uniaxial tensile behavior of the Ti-6Al-4V indicates that H (at least up to 500 wt. ppm) has no significant effect on the following: yield

stress, strain-rate hardening exponent, strain hardening exponent and plastic anisotropy. In addition, the fracture limit diagram for the Ti-6Al-4V sheet indicates that (like Ti-30V) there is no significant influence of hydrogen on the fracture behavior over a range of stress states from uniaxial tension to balanced biaxial tension for the equiaxed α/β microstructure examined.

In this study[70] the surprising result is that even in a test as severe as equibiaxial tension, both the Ti-30V and Ti-6Al-4V sheet remain immune to hydrogen embrittlement even at high levels of hydrogen in solution, up to $\sim$ 9 wt.% in the case of the β-phase alloy. We thus conclude that H is not likely to cause hydrogen embrittlement in any Ti alloy if hydrides do not form and provided that phase stability is retained. Given that the β-phase appears to act as a relatively innocuous sink for H, the resistance to hydrogen embrittlement of an α-phase alloy such as CP Ti can be substantially improved by adding a small amount of an element which stabilizes the β-phase and causes formation of at least a small volume fracture of the β-phase upon heat treatment.

6. The Deformation and Fracture of Alloys Containing Pores and Voids (Roy Bourcier, Ph.D., Paul Magnusen, Ph.D. candidate, Ellen Dubensky, M.S. candidate, Drs. Owen Richmond and Ron Smelser, Alcoa Laboratories)

Titanium-base alloys have been used extensively in applications requiring high strength to weight ratios, good toughness, and resistance to environmental degradation. Despite their attractive properties, the high cost of producing components of complex shape often precludes the use of Ti alloys. This situation has led to a demand for the application of advanced processing techniques (such as powder metallurgy, P/M) to Ti alloys and, for similar

reasons, to other high performance alloys as well. Inherent in these processing methods is the possibility that the resultant components may contain defects, porosity for example, not normally present in a cast and wrought components. Since these defects can seriously degrade certain properties, such as resistance to fracture, it is necessary to either eliminate the defects or at least to be able to predict accurately the defect-induced failure conditions.

The effects of pre-existing porosity, matrix strain hardening, and strain rate on the deformation and fracture of high strength engineering alloys has been examined using primarily PM Ti alloys as model systems[71,72]. The experimental aspects of the study have been based on the contrasting deformation and fracture behavior of two Ti alloys (commercially pure (CP) Ti and Ti-6Al-4V) and pure Ni, each of which possess considerably different strain hardening characteristics. The materials have been consolidated via powder metallurgy techniques to similar levels of rounded porosity and examined on the basis of the yielding, flow, and void growth behavior. The resulting behavior has been analyzed on two levels: (1) a bulk porosity basis simulated by a large strain elastic-plastic finite element model and (2) a local porosity basis in which the material is viewed in terms of planes of high pore content: "imperfections". The principal results may be summarized as follows:

1. Increasing porosity causes decreases in the (a) yield stresses (which exceed those predicted by the rule of mixtures), (b) ductility, and (c) strain hardening exponent. In all three cases, the decreases are most pronounced in the material [Ti-6Al-4V] with the smallest work hardening rate. Over the range of 10^{-4} to 10^{+2} s^{-1}, strain rate has little if any influence on the above behavior.

2. At all porosity levels, the fracture-surface is characterized by a much higher pore content (roughly from 4x to 10x, depending on material) than is present on a random plane in the bulk. The large amount of porosity on the fracture surface cannot be accounted for by strain-induced pore growth.

The decrease in tensile ductility with increasing porosity is both pronounced and technologically important. There appear to be two mechanisms which cause the loss of ductility: (1) a decrease in uniform elongation due to the combination of pore growth and a decrease in work hardening rate and (2) porosity-triggered shear instabilities which occur at large strains and are caused by planes of high pore content ("imperfections").

The above studies indicate that porosity (or strain-induced voids) introduce planes of weakness (or "imperfections") which subsequently trigger a shear instability creating microvoid sheets and final failure. Obviously, the distribution of pores/voids (or other processing "defects" such as non-uniform microstructures) should be an important and often controlling parameter in such a fracture process. Thus, in a unique study we are also modeling the influence of void/pore distributions on ductile fracture using two dimensional arrays of holes whose positions are predicted by a computer on the basis of random-numbers[73]. Initially, the experiments are being conducted on arrays of equi-sized holes in which the (a) area fraction of holes, (b) diameter of the holes, and (c) minimum spacing of the holes is controlled. The study is based on two materials (1100 Al and 7075 Al) of differing work hardening rates which are tested under conditions of plane stress vs. plane strain deformation between the holes. The following preliminary conclusions may be made for the area fraction of holes ranging from 2.5 to 5.0%: (1) for plane-stress fracture of a material with a moderately high work hardening rate, the tensile

ductility and yield stress are most strongly dependent on hole diameter, and (2) for plane-strain fracture in a material with comparatively low work hardening, hole spacing dominates ductility while yield stress is most sensitive to the area fraction of holes. This research is presently being extended to 7075 Al sheet, and the analysis will be extended to include a wider range of hole parameters. Implications to the theory of ductile fracture will also be drawn.

As the above studies show, ductile fracture in high-strength alloys is a result of the termination of stable plastic flow by catastrophic strain localization. Microstructurally, this flow localization can often be traced to the presence of large voids within the material. A large body of work has been performed in recent years in an attempt to identify the mechanical and microstructural variables that determine the deformation and fracture behavior of a material containing voids. However, while considerable theoretical work has modeled voids as holes in a regular (non-random), two-dimensional array, no study has yet been performed in which the flow and fracture behavior near an individual hole or a pair of holes is examined both experimentally and analytically. Thus, the tensile behavior of plane-strain specimens each having a central hole with axis in the zero-strain direction has been examined[74]. The study was based on the contrasting behavior of two materials, one with a relatively high strain hardening rate (an HSLA steel) and the other with a low rate (Ti-6Al-4V). Deformation of the holes, associated necking of the ligaments, as well as the overall force-elongation response exhibit excellent agreement with prediction from a large-strain elastoplastic finite-element model. Failure of the high strain-hardening material occurs by ductile tearing across the ligaments, whereas failure of the low-hardening material occurs by shear localization. This is consistent with the predicted

incremental plastic strain distributions as calculated by the finite element method. The experimental results and predictions of the finite-element models indicate the importance of work hardening in diffusing plastic flow in the presence of a geometric inhomogeneity.

In a final, related study, the deformation and fracture between pairs of holes has been studied in a 7075 Al alloy[75]. The test technique, which is unique, utilizes in a final, related study, tensile samples each with a pair of holes through the thickness. The holes are sufficiently close (two hole diameters apart) so as to concentrate slip between them, and the local states of stress and strain can be controlled by the orientation of the holes to the stress axis. Fracture occurs by flow localization between the holes and very little hole growth occurs, especially in the T6 condition. Fractography indicates a transition in fracture appearance with hole orientation with dimpled fracture predominating, but the degree of shear fracture increases as the pair of holes becomes more inclined to the stress axis. A modification of the Bridgeman analysis is used as a description of the approximate state of stress between the holes as a function of the hole orientation. The results indicate that the criterion for ductile fracture depends on both the state of stress and the strain state, but the functional dependence could not be determined. The results strongly suggest the importance of flow instability in the void link-up process, especially in plane-strain conditions.

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